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AGEING OF FORGED ALUMINIUM-LITHIUM 8091 ALLOY

by

P. D. Pitcher

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AGEING OF FORGED ALUMINIUM-LITHIUM 8091 ALLOY

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SUMMARY

The microstructures obtained in forged 8091 alloy after using selected ageing heat treatments have been studied by transmission electron microscopy, and the corresponding tensile and fracture toughness properties measured. It was found that ageing by heating at a controlled rate of 10°C/h from room temperature to the ageing temperature (170°C or 190°C), and then holding at this temperature, led to a finer, more uniform distribution of the S' phase compared to normal direct ageing. This improved the strength/toughness balance of the alloy.

This work has previously been published in Scripta Metallurgica, 22, p 1301 (1988).

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1 INTRODUCTION

As part of an extensive research programme on the development of forgings made in the aluminium-lithium alloy 8091, efforts have been made to optimise the ageing heat treatments applied to this alloy. In plate and sheet products, the use of cold work by stretching prior to ageing, combined with the use of an ageing temperature of 150°C to 170°C has provided a method of obtaining good toughness at high strength levels in both 8090 and 8091 alloys[1]. The dislocations introduced by the cold work provide sites for the uniform heterogeneous precipitation of S' phase (Al_2CuMg), while the use of ageing temperatures of 170°C or less avoids excessive grain boundary precipitation and improves toughness. However, although the application of cold work to forgings by cold compression has a similar effect in enhancing S' phase precipitation, it is not always feasible to achieve a uniform degree of cold compression, especially in certain die forgings. An alternative means of optimising the balance between strength and toughness is therefore desirable. This Memorandum describes work on the microstructures and properties obtained in a forged 8091 alloy following ageing by heating at a controlled rate of 10°C/h from room temperature to the ageing temperature, and holding at this temperature for selected times.

2 EXPERIMENTAL PROCEDURE

The 8091 alloy was supplied as ingot by Alcan International Limited. The chemical composition of the alloy was as follows (weight %): Al-2.40Li-2.07Cu-0.76Mg-0.11Zr. Simple hand-forged slabs (approximate size 250 mm × 170 mm × 80 mm) were produced from blocks of the ingot, the forging temperature being 490°C. The blocks were solution heat treated at 545°C and quenched into polyalkalene glycol at about 35°C. Ageing was carried out by heating from room temperature to temperatures of 150°C, 170°C and 190°C at a rate of 10°C/h and holding at temperature for selected times. Microstructural observations were first made using small samples; larger pieces from the forgings were later given selected heat treatments and the tensile and toughness properties measured using standard one-quarter inch BSF tensile and 20 mm thick compact tension fracture toughness test pieces. The microstructures and properties were compared to those obtained by normal direct ageing at either 170°C for 48 hours or 190°C for 16 hours. Thin foils for transmission electron microscopy (TEM) were prepared using jet polishing equipment with 30% HNO_3 in methanol at a temperature of approximately -20°C as the electrolyte. TEM was carried out using a JEOL 200CX operating at 200 kV.

3 RESULTS

The typical microstructures obtained on direct ageing at 170°C/48 hours and 190°C/16 hours are shown in Fig 1. The essential features are a uniform dispersion of the fine δ' (Al_3Li) precipitates with a narrow precipitate free zone (pfz) at grain boundaries (about 60-80 nm wide), heterogeneous precipitation of S' phase laths on dislocations and subgrain boundaries, a relatively wide pfz (about 0.5 μm) for the S' phase in regions of more widespread precipitation of S' phase, and many particles on high angle grain boundaries (mainly of the δ phase (AlLi), the icosahedral phase (Al_6CuLi_3) and iron-containing phases[2]). Similar features were observed on slow heating to the ageing temperature, although there were also some significant differences.

Heating at 10°C/h to 150°C and holding for 88 hours resulted in much less grain boundary precipitation and mainly heterogeneous precipitation of S' phase laths which often formed in quite large loops of up to 0.5 μm in diameter, Fig 2. Ramping to 170°C and ageing for 21 hours again resulted in quite clean grain boundaries and mainly heterogeneous S' phase, although a little homogeneously nucleated S' could also be seen in the matrix, Fig 3. Finally, heating slowly to 190°C and ageing for 19 hours led to an increase in grain boundary precipitation, but also to a more uniform and denser distribution of fine, homogeneous S' phase, as seen in Fig 4.

Following such observations, larger blocks of the forged slabs were aged for either (a) 10°C/h to 170°C and hold for 48 hours, or (b) 10°C/h to 190°C and hold for 10 hours. The tensile and fracture toughness data obtained are given in Table 1, and compared to values for direct ageing of the same material at 190°C/16 hours.

The slow heating to 190°C plus 190°C/10 hours has improved strength values compared to the direct ageing 190°C/16 hours, although the fracture toughness values are very similar. Short transverse tensile elongation and toughness values are still too low at the higher strength levels. However, oxides and inclusions related to flux or grain refining elements have been observed on some short transverse fracture surfaces, which would at least partly account for the low elongation values. Ramping the temperature slowly to 170°C and holding for 48 hours resulted in rather lower strength levels compared to the slow heating to 190°C, especially for the proof stress, but better tensile ductility and fracture toughness. It should be noted that the longitudinal tensile test pieces did not have a very strong longitudinal grain flow, and the longitudinal tensile elongation values were therefore lower than expected.

The slow heating to the ageing temperature enhanced the formation of fine homogeneous S' phase in the forgings in the same way as it did in the smaller samples, Fig 5.

4 DISCUSSION

It has been shown that slow heating to the ageing temperature can lead to a better balance of tensile strength and toughness, *ie* similar or slightly improved toughness values are achieved at a higher strength level. The increase in strength obtained by slow heating to 190°C must at least partly be due to the additional time at elevated temperature before the ageing temperature is reached, and the δ' precipitates are larger for a given time at the actual ageing temperature. For example, measurements of precipitate diameters for TEM dark-field images of the δ' precipitates gave average values of 25 nm for the slow heating to 190°C and holding 10 hours, and 20 nm for direct ageing at 190°C/16 hours. This means that for a given required δ' precipitate size (strength level), a shorter time at the ageing temperature is needed. It was hoped that this would reduce the amount of grain boundary precipitation, but evidence for this was inconclusive. An additional benefit of the slow ramping to the ageing temperature is the enhanced precipitation of homogeneous S' phase. S' phase is believed to benefit mechanical properties by promoting cross-slip and reducing the tendency for planar deformation in Al-Li-Cu-Mg alloys (*eg* Refs 3 and 4). This improves ductility, and hence the UTS, although the proof stress appears to be little affected. Ramping to the ageing temperature may therefore not benefit proof stress significantly, as indicated by the lower values of proof stress after slow heating to 170°C and ageing for 48 hours, and is certainly not as effective as cold working in this respect.

The mechanism of enhanced precipitation of S' phase by using slow heating to the ageing temperature is probably related to the release of vacancies. It has been suggested elsewhere that the vacancies which are trapped in solid solution on quenching (being strongly bound to lithium atoms) are released during natural ageing as the δ' precipitates grow[5], although other work has indicated that the natural ageing response in 8090 and 8091 alloys is due to an S' phase precursor rather than the growth of δ'[6]. In the present case the δ' precipitates will grow as the temperature is raised during the slow heating to the ageing temperature, and vacancies will be released. It has been reported elsewhere that dislocation loops are formed around particles of Al_3Zr in the quenched condition, and that there is enhanced nucleation of individual δ' precipitates on the Al_3Zr particles[7]. It was suggested that the excess vacancies around Al_3Zr particles, due to the enhanced δ' precipitation, condense to form the loops which can expand on ageing and provide sites for subsequent nucleation of S' laths, as observed in Figs 1 to 3. It should be noted that Al_3Zr particles are not observed at the centre of all such loops, and are therefore not essential for their formation. The slow heating to the ageing temperature perhaps results in a more controlled release of the vacancies as the δ' precipitates coarsen, (compared to that for direct ageing), and as ageing progresses, homogeneous S' is also formed more readily as extra vacancies which have not condensed to form loops become available; *ie* at a temperature high enough for precipitation of S' phase, the vacancies assist the formation of homogeneous S' phase rather than condense to form loops, perhaps by aiding the widespread formation of an S' phase precursor.

The enhanced precipitation of fine, homogeneous S' phase may improve resistance to stress corrosion cracking (SCC). It has been shown that cold work prior to ageing decreases SCC susceptibility, and it is believed that this is due to the precipitation of S' phase on the dislocations within grains; this reduces the driving force for preferential anodic dissolution at grain boundaries where there are copper-rich particles[8]. It may be expected that anisotropy effects in unrecrystallised grain structures will also be reduced by extensive matrix precipitation of S' phase.

5 CONCLUSIONS

Ageing of forged 8091 alloy by heating at a controlled rate of $10^{\circ}C/h$ from room temperature to the ageing temperature ($170^{\circ}C$ or $190^{\circ}C$), and then holding at this temperature, leads to a finer, more uniform distribution of the S' phase compared to normal direct ageing. This improves the strength/toughness balance of the alloy, and could be useful where cold work prior to ageing is not possible.

Table 1

TENSILE AND FRACTURE TOUGHNESS DATA

	0.2% proof stress (MPa)	UTS (MPa)	Elongation (%)	K_Q (MPa $\sqrt{m}$)
10°C/h to 170°C + 170°C/48 h	L 412	523	10	L-LT 23*
	ST 368	425	4	ST-L 13
10°C/h to 190°C + 190°C/10 h	L 454	528	5	L-ST 22
	ST 407	450	0.5	ST-L 10
190°C/16 h	L 415	489	4	L-LT 22
	ST 385	413	0.5	ST-L 11*

NOTE: Tensile data are the mean of two tests.
Starred values are valid K_{IC} results.

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(a)

(b)

(a) 170^{o}C/48 h

(b) 190^{o}C/16 h

Fig [illegible] Microstructure after direct ageing

Fig 2

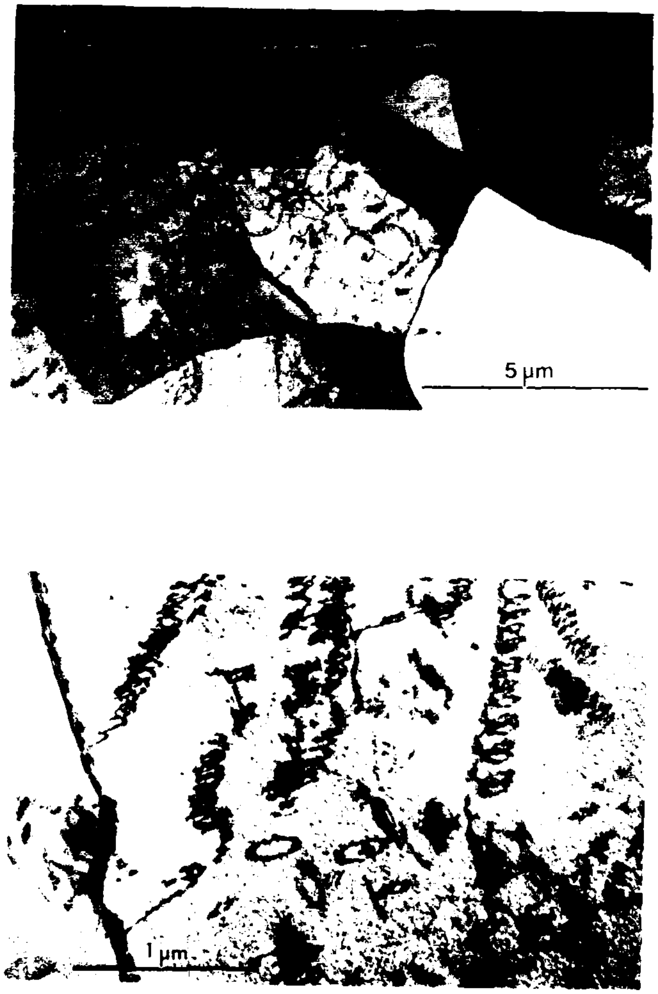

Fig 2 Microstructure after heating at 10°C/h to 150°C and holding for 88 h

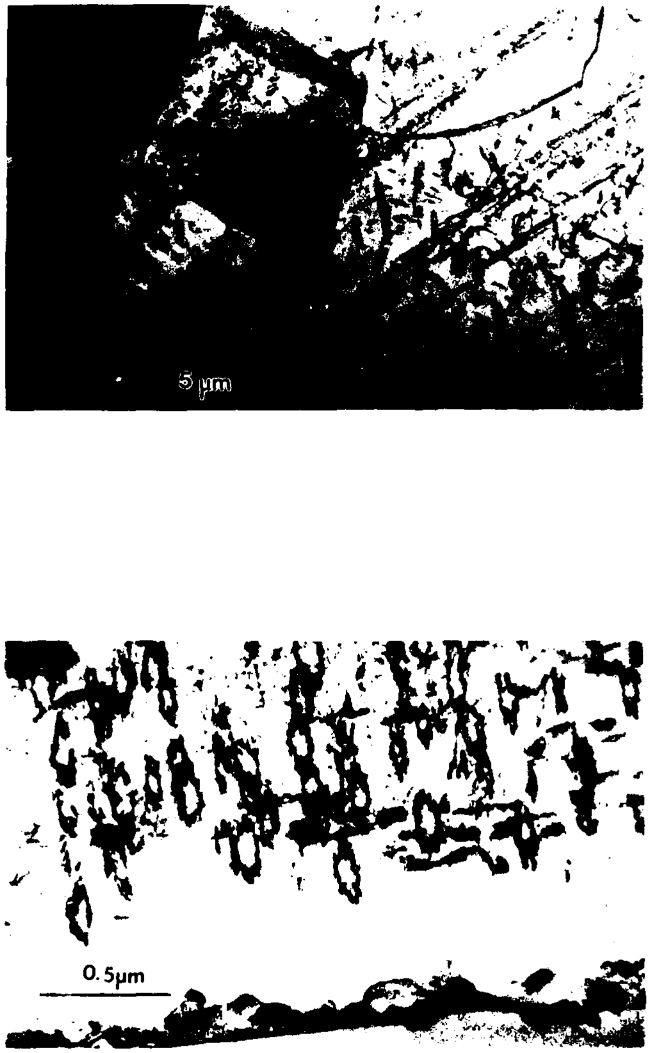

Fig 3 Microstructure after heating at 10^{o}C/h to 170^{o}C and holding for 21 h

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Fig 4

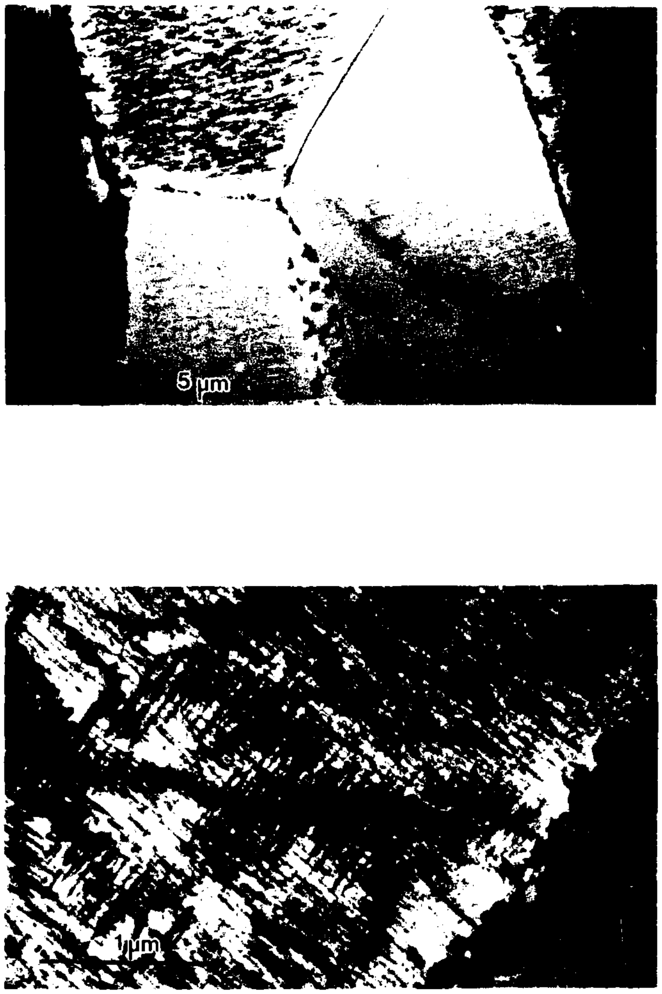
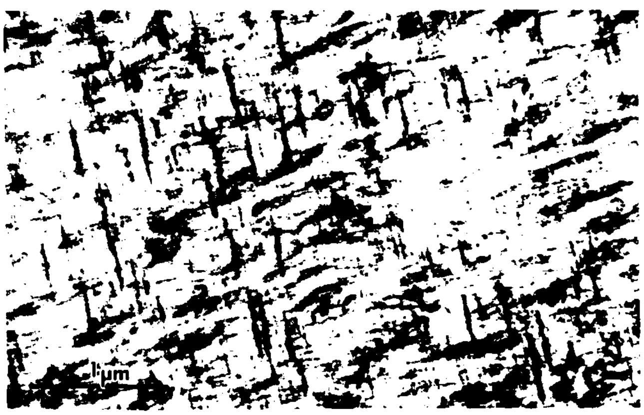


Fig 4 Microstructure after heating at $10^{o}C/h$ to $190^{o}C$ and holding for 19 h

(a)



(b)

(a) $10^{o}C/h$ to $170^{o}C + 170^{o}C/48$ h

(b) $10^{o}C/h$ to $190^{o}C + 190^{o}C/10$ h

Fig 5 Microstructures in forged blocks

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17. Abstract

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